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TEXTURE STRENGTHENING OF TITANIUM AND ZIRCONIUM (TEKSTURNOE UPR--ETC(U)
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**TEXTURE STRENGTHENING
OF TITANIUM AND ZIRCONIUM**

by
N. V. Ageev
A. A. Babareko

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(6) TEXTURE STRENGTHENING OF TITANIUM AND ZIRCONIUM
(TEKSTURNOE UPROCHNENIE TITANA I TSIRKONIYA)

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A. A. Babareko

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EDITOR'S SUMMARY

→ ^{alpha} The development of crystallographic textures and the corresponding effects on strength and hardness in Ti, Zr, and their alloys were studied. The addition of α or β stabilizing elements to Ti, for example, changed the tendency towards dislocation formation, and encouraged the development of the base texture. By varying the temperature, the deformation rate, and the amount of reduction during rolling, in addition to the orientation of the principal deforming forces, the type of crystallographic texture developed in α and ($\alpha + \beta$) Ti alloys assumed various predictable forms, many of which gave special mechanical properties to parts made from these alloys. Analogous effects were obtained for Zr. ↙

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The yield point of metallic crystals depends on the crystallographic orientation to the stress axis. It is associated with discrete plastic flow along planes and in directions corresponding to the close-packing of atoms. The nature of the orientation distribution in shear systems of crystals of different symmetry is explained by this dependence of yield point.

In metals with a cubic lattice, such as iron, nickel, and copper, variations in yield point are due to change of the Schmid orientation factor for the twelve slip systems, $\{1\bar{1}0\}\langle 111 \rangle$, in a body-centred lattice, and $\{111\}\langle 110 \rangle$ in a face-centred lattice. The value of the Schmid factor $\cos \phi \cos \theta$ defines the applied stress σ for initiation of flow of the material, characterised by some value of the critical shear stress τ ¹

$$\sigma = \tau / \cos \phi \cos \theta$$

where ϕ and θ are the angles of the normal to the shear plane, and the direction of shear with respect to the applied stress axis, respectively. In cubic crystals the value of the Schmid factor for a primary shear system varies from 0.5, for a crystal with orientation of the stress axis in the centre of the stereographic triangle, to 0.27 in a crystal with $\langle 111 \rangle$ axial orientation. Variations of the Schmid factor define the limits of possible orientation strengthening and show that the maximum difference in yield point for a cubic crystal can be as high as a factor of 1.8.

For hexagonal, as opposed to cubic metals, there is typically greater irregularity in the distribution of lattice orientation. In heteropolar crystals, three slip systems, $(0001)\langle 11\bar{2}0 \rangle$ and/or six systems, $\{10\bar{1}0\}\langle 11\bar{2}0 \rangle$, act with lowest critical shear stresses in the close-packed $\langle 11\bar{2}0 \rangle$ directions. All the $\langle 11\bar{2}0 \rangle$ directions lie in a single crystallographic plane, (0001) , and the Schmid factors for slip vary from 0.5 to 0. In hexagonal crystals, there are orientation domains where the Schmid factors for primary slip have values close to zero. Twinning and slip along non close-packed directions, with critical shear stresses greater than for $\langle 11\bar{2}0 \rangle$ shear, are additional deformation mechanisms in hexagonal crystals. This leads to marked orientational dependence of the yield point for heteropolar crystals. Thus, in pure beryllium, the initial shear stresses for $(0001)\langle 11\bar{2}0 \rangle$ and $\{\bar{1}212\}\langle 11\bar{2}3 \rangle$ slip differ by a factor of 10^3 , giving, respectively, values of 0.22 and 200kg/mm^2 , and the three-dimensional plasticity of this metal is problematical². For titanium, as a result of the development of twinning, the ratio of the critical shear stresses

$\tau\{10\bar{1}0\}/\tau_{\text{twinning}}\langle 11\bar{2}0\rangle$ amounts to $\approx 1:30$ and the yield points of titanium crystals, under compression at room temperature along the c , $\langle 0001\rangle$ axis, and in a direction perpendicular to this, differ by a factor of less than 10 for a stress orientation accuracy of 0.5° ³.

In a polycrystalline metal, the creation of preferential crystal orientation and texture permits use of the orientation dependence of yield point for enhancing its strength in comparison with that of the untextured material. Texture strengthening is a function of yield point orientation dependence, and the type and degree of texture perfection of the material. It also depends on the mode of stressing.

In a polycrystalline material it is difficult to create a texture approaching the degree of perfection in a single crystal. Texture strengthening therefore comprises only an insignificant fraction of the theoretical value calculated from the properties of single crystals. It has been shown experimentally³ for titanium single crystals that increasing the disorientation of the stress axis from 0.5° to 2° sharply decreases the yield point orientation dependence, $\sigma_{\tau,\parallel c}/\sigma_{\tau,\perp c}$ from 10 to 6. The range of scatter due to crystal orientation in a textured polycrystal usually fluctuates from several degrees to tens of degrees. This limits the extent to which the theoretically possible value for texture strengthening can be attained.

As a result, however, of strongly developed anisotropic yield in titanium and zirconium, texture strengthening is successfully used to enhance the structural strength of alloy components functioning under conditions of biaxial tension^{4,5}. It has been shown that, for a balanced biaxial stress with a stress ratio of 1:1 (high pressure spherical vessels), the $\langle 0001\rangle$ texture is effective in the developed spherical surface. For an unbalanced stress with a stress ratio of 1:2 (high pressure cylindrical vessels), the $\{hki0\}\langle HK10\rangle$ texture is similarly useful in the developed cylindrical surface, with orientation of the normals to the prism planes in the radial and axial directions, and normals to the basal plane in the tangential direction. Under such stresses, the presence of the above textures leads to a deficiency of easy-slip systems, because of the zero or low values of the reduced shear stresses in the directions of closest packing of atoms $\langle 11\bar{2}0\rangle$.

On account of high resistance to thinning in the $\langle 0001\rangle$ direction, a twofold increase in the strength of sheets of zirconium and its alloys having the $\langle 0001\rangle$ texture has been obtained during compression across the thickness

(small-scale operation of biaxial balanced stretching with reduction of hydrostatic stress components)⁶. By creating an α -phase texture close to the ideal basal texture, by balanced biaxial stretching of ($\alpha + \beta$) Ti-6Al-4V alloy, a strength increase of 41%, in comparison with the uniaxial properties of the sheet, has been obtained⁷. In the α -alloy, Ti-4Al, with similar type of texture and a polar density of 2-4, 64% texture strengthening has been achieved. Other values were 34% for Ti-5Al-2.5Sn; 26% for commercial titanium with double basal texture, strengthened by oxygen; 38% for soft 1mm titanium sheet and 44% for 3.5mm sheet⁸. In domestic α -alloy of VT-5-1 composition, approximating to Ti-5Al-2.5Sn, 45% texture strengthening was obtained⁹ by enhancing the perfection of the central-type basal texture (polar density 5-7 units).

For VT-14 ($\alpha + \beta$) alloy with multicomponent prismatic + double basal texture, it has been shown that reinforcement of the basal component in the texture of an annealed sheet leads to insignificant texture strengthening; $k = \sigma_{\text{biaxial}} / \sigma_{\text{uniaxial}} = 1.15$. However, for sheets subjected to thermal strengthening treatment, with decreased plasticity, the texture becomes of paramount importance. In sheets of this sort, the coefficient of texture strengthening under biaxial balanced stress varies from 0.77 in sheet with a prismatic texture, to 1.12 in sheet with a deflected basal texture $\{10\bar{1}3\}$, with a density of ≈ 6 polar units. The presence of the latter obviates texture weakening of thermally strengthened VT-14 alloy, with the exception of basal slip under the action of tensile stresses in the plane of the sheet, and gives the possibility of realising the results of strengthening treatment under rigid biaxial load conditions⁹. Thus, in hexagonal alloys, with somewhat reduced plasticity due to the presence of internal stresses associated with phase cold-hardening (alloys strengthened by heat treatment), or due to the presence of brittle phases, two aspects of the orientation dependence of deformation properties arise: (1) for enhancing the plastic properties of such alloys under compound rigid load they must have a texture obviating brittle slip along the (0001) cleavage plane; (2) it is desirable that this texture should contribute towards strengthening by virtue of a deficiency of easy slip.

In industrial alloys of titanium and zirconium a very damaging effect on plasticity is caused by the precipitation of a hydride phase. In titanium alloys precipitation of this phase is accompanied by strong local phase cold-hardening, due to a 20% disparity between the specific volume of the matrix and that of the hydrides. High internal microstresses in the zones of induced

growth can serve as a source of local crack initiation¹⁰. The cleaving action of hydrides and crack formation have been observed¹¹ in Ti-5Al-2.5Sn alloy during investigation under stress. Under compound loading, orientation of the hydrides has a marked effect on properties. This orientation depends, in turn, on the texture of the material. In tubes of zircalloy-2 alloy with radial orientation of hydride flakes, a loss of texture strengthening of 40% has been observed, in comparison with a control hydride-free alloy. With tangential hydride orientation, however, the loss was only 20% (experiments using unbalanced biaxial tension)¹²⁻¹⁴.

Empirical and semi-empirical yield curves have been constructed^{15,16} showing texture strengthening and weakening characteristics under various experimental conditions for tubes of zircalloy-2 and -4 with different textures and hydride content and distribution. These curves show the variation of relative yield point under biaxial loading as a function of the value of α , the ratio of the applied stresses (Fig.1). In the presence of hydrides the shape of the curve is distorted, being shortened in the directions for $\alpha = 1:1$ and $1:2$. This indicates partial or total loss of texture strengthening in a tube subjected to biaxial stress. Greatest distortion was observed for zircalloy-4 containing hydrides with the texture of the tube having the $\langle 0001 \rangle$ direction in the tangential direction, i.e. with a texture theoretically most favourable for enhancing the structural properties of the tube under biaxial unbalanced stress. For tubes containing hydrides and having a double basal or basal texture in the radial direction, the shape of the yield curves is less distorted. This suggests that the theoretically most favourable texture for structural components is not always effected. Thus, for tubes with the $\langle 0001 \rangle$ texture in the tangential direction and with precipitation of hydrides, it is possible that the mechanism of brittle failure along the basal plane normal to the direction of the maximum applied tensile stresses may play a part. Theoretically, less advantageous tube textures, basal and double basal in this case, may give a more marked contribution to texture strengthening.

The property data given for titanium and zirconium alloys indicate the importance of studying texture formation in these alloys. In polycrystalline aggregates the texture develops during directional crystallisation, plastic deformation and recrystallisation of the pressure-treated material and also during orientated phase transformation. In a metal with polymorphic transformations, thermal treatment is accompanied by epitaxial multivariant transitions

from one structure to another. This leads to complication of the original texture. Thermal cycling through the polymorphic transition temperature is used to weaken the original texture. In a number of cases, however, inheritance of the initial orientation is observed. This requires explanation.

In titanium and its alloys the various types of texture due to pressure treatment depend on differences in the crystallographic deformation mechanism of the α -phase, on the conditions of treatment, and on the composition. Strongly cold-worked pure titanium with well developed deformation twinning and prismatic slip forms a double basal texture, divergent from the rolling plane by 25° , with traces of prismatic orientations. Alloys have a strong influence on the deformation mechanism and structure of sheets. Impurities in titanium produced by thermal reduction with magnesium favour the development of a useful texture of a central basal type in sheets, but, in hydride-containing metal, textures with increased basal density in the region of the elliptical ring round the normal to the rolling plane¹⁷ occur. Suppression of twinning during an increase of rolling temperature or a decrease of the rate and degree of reduction leads to the development of a central basal type texture in titanium.

Alloying titanium with β -stabilisers increases its tendency to twinning and the development of a divergent basal texture under any conditions of temperature and reduction^{18,9}. Alloying with α -phase stabilising elements favours the development of a central-type basal texture. A study has been made of texture development in sheets of VT-5-1 alloy containing aluminium and tin. Continuous growth of the texture intensity with increase in the degree of reduction at 800°C was shown. Reductions of more than 70% led to a basal polar density of 5-7 units in the plane of the sheet. A cold deformation of 30% did not destroy the central-type basal texture⁹. In this alloy, in comparison with pure titanium, there was an increase in the density of packing deformation defects in the basal plane²⁰. Evidently, the splitting of basal dislocations somewhat reduces the critical basal plane stresses, increases the probability of basal slip in comparison with prismatic slip and twinning, and favours the development of a basal type texture under any conditions of temperature and reduction.

Recrystallisation after pressure treatment at temperatures below that of the phase transition has little effect on the type of texture of α -titanium, the texture only varying in its degree of scatter^{9,18,19}. At the same time, reheating above the transition temperature can radically modify the texture of

the α -phase. This happens due to multivariant orientation transformations, $\alpha \rightarrow \beta \rightarrow \alpha$, along many β -phase habit planes, $\{110\}_{\beta} \parallel (0001)_{\alpha}$, and involves complication of the initial α -phase texture. Imposition of another process is also possible, i.e. prolonged soaking in the β -phase range leads to a recrystallised texture in this phase, different to the texture at transformation. As a result of the reverse transformation, $\beta \rightarrow \alpha$, new types of texture can appear in the α -phase, not characteristic of this phase after pressure treatment or recrystallisation (Fig.2)⁹.

Industrial processes for the manufacture of titanium alloy stock materials include pressure treatment and heating at temperatures corresponding to different phase states. Phase transformations, with or without load, have a strong influence on the texture of the end product. Studies of the conditions for texture formation in complex industrial processes are very important in connection with the effect of alloy texture on the mechanical properties of half-finished materials. Interpretation of texture formation mechanisms offers the possibility of directing industrial processes towards obtaining materials with given useful textures and taking advantage of the reserve of strength present in a heteropolar phase.

Planar deformation of $(\alpha + \beta)$ titanium alloys in the temperature range for existence of the β -phase, at the start of the pressure treatment, with completion of deformation in the $(\alpha + \beta)$ range, leads to the formation of a new type of fine single or two component transformation texture, not previously described for the α -phase of titanium. The new texture is strongly characterised by a prismatic component and an associated pyramidal orientation^{9,21}. In contrast to thermal cycling without load, which complicates and diffuses the texture of the α -phase²², deformation for several cycles with repeated transitions through the $\beta \rightleftharpoons \alpha + \beta$ polymorphic transformation point does not lead to a textureless alloy state. High-temperature stamping of VT-3-1 alloy at the start of deformation, at temperatures up to 1000°C, with completion in the $(\alpha + \beta)$ range, leads to the development of a $\{11\bar{2}0\} + \{10\bar{1}2\}$ texture with a fractional ratio, $f_{11\bar{2}0} : f_{10\bar{1}2}$ of 2:4 or higher. Deformation of VT-14 at lower temperatures, close to the limit of the $\beta \rightarrow (\alpha + \beta)$ transformation, is accompanied by formation of a $\{11\bar{2}0\} + \{11\bar{2}4\}$ texture with a 1:1 component ratio (Fig.3). The textures of the residual β -phase in these alloys are different, having orientations of $\{100\}$ in VT-3-1 and $\{111\}$ in VT-14 alloy. Theoretical polar diagrams for the transformed α -phase, in association with initial β -phase, are constructed according to Burgers, giving an epitaxy of close-packed planes

and directions in the two lattices $(0001)_\alpha \parallel \{110\}_\beta$ and $\langle 11\bar{2}0 \rangle_\alpha \parallel \langle 111 \rangle_\beta$. Fig.4 shows theoretical direct polar diagrams (0001), characterising the texture of the β -phase, transformed along all possible habit planes of the α -phase, with planar deformation textures (a) $\{111\}\langle 1\bar{1}0 \rangle$, (b) $\{100\}\langle 110 \rangle$ and (c) recrystallisation textures $\{110\}\langle 001 \rangle$ and $\{110\}\langle 1\bar{1}0 \rangle$. The theoretical texture of the transformed α -phase for an initial texture of the β -phase, $\{111\}\langle 1\bar{1}0 \rangle$, is characterised by the presence of $\{11\bar{2}0\} + \{11\bar{2}4\}$ components in the ratio 1:1, this being also experimentally observed for VT-14 alloy with a texture of the residual β -phase $\{111\}$. During orientation of the initial β -phase, $\{100\}\langle 110 \rangle$, the theoretical texture of the transformed α -phase is characterised by $\{11\bar{2}0\} + \{1\bar{1}02\}$ components in a ratio of 1:2. For VT-3-1 alloy with a residual β -phase texture $\{100\}$, multi-variant transformation theory correctly predicts only the crystallography of the transformed texture components under α -phase loading and does not explain the experimentally observed high intensity ratio $\{11\bar{2}0\}:\{1\bar{1}02\} \approx 4$, or the formation, in a number of cases, of even the single component texture $\{11\bar{2}0\}$ ²¹.

A study has been made of the effect of load on the mechanism of the $\beta \rightarrow \alpha$ transformation. In accordance with Le Chatelier's principle and in the presence of externally applied stresses, the transformation process will have a high rate along that one of the possible six habit planes $\{110\}$ where stress relaxation is maximal. Epitaxial $\beta \rightarrow \alpha$ transformation, arising from the lattice parameters of the two phases, requires elastic tensile deformation in the $\langle 11\bar{0}0 \rangle_\alpha \parallel \langle 1\bar{1}0 \rangle_\beta$ direction, compression along $\langle 11\bar{2}0 \rangle_\alpha \parallel \langle 110 \rangle_\beta$ in the habit plane, and extension in the $\langle 0001 \rangle_\alpha \parallel \langle 110 \rangle_\beta$ direction perpendicular to the habit plane. For an initial orientation of the β -phase, $\{100\}\langle 110 \rangle$, in the sheet, or during stamping, $\beta \rightarrow \alpha$ transformation along habit planes perpendicular to the direction of stretching will satisfy the condition for maximal stress relaxation. During stamping of VT-3-1 alloy with a $\{100\}$ β -phase texture, the habit planes normal to the stretching direction condition development of a single component $\{11\bar{2}0\}$ texture in the transformed α -phase. Hardening the alloy and ageing in the absence of externally applied stresses increase the component with $\{1\bar{1}02\}$ texture, the fractional ratio $\{11\bar{2}0\}:\{1\bar{1}02\}$ being close to the theoretical value for multi-variant transformation²¹. The formation of a (110) β -phase recrystallisation texture during high-temperature pressure treatment of titanium alloys and subsequent hardening can lead to $\beta \rightarrow \alpha$ transformation along the (110) habit plane lying in the plane of the sheet or stamping, and to development of an α -phase basal texture (see Fig.4c). Hardening stresses of opposite sign to those applied during deformation can create conditions for the preferred action of this habit plane.

With an initial $\{111\}\langle 1\bar{1}0 \rangle$ β -phase orientation, deformation during rolling does not create significant advantages for transformation along any one of the habit planes, all of which are more or less equivalent, and conditions for maximal stress relaxation are not fully satisfied. Transformation of β -phase with such a texture, during hot rolling under load, follows a multivariant mechanism with formation of a $\{11\bar{2}0\} + \{11\bar{2}4\}$ texture and with the theoretical ratio of the components. Subsequent rolling of VT-14 alloy in the $(\alpha + \beta)$ state at 800°C introduces a component with a divergent basal deformation texture of the β -phase, $\{10\bar{1}2\}$, and weakens the prismatic component of the transformed α -phase texture (Fig.5a). A non-monotonous relationship has been observed between the value of the useful basal fraction and the degree of alloy reduction, a maximum being seen at deformations of $\approx 70\%$ ⁹. Lowering the rolling temperature to $700\text{--}600^\circ\text{C}$ stabilises the prismatic component of the texture, evidently due to developed prismatic slip at these temperatures. The initial prismatic orientation formed as a texture transformation component during hot rolling, is, in the case of developed prismatic slip during subsequent warm rolling, the final stable orientation and does not undergo any changes. In sheets used as stock for pressing spherical vessels, the preservation of this textural component is undesirable. Cross-rolling of stock at specified temperatures avoids prismatic texture stabilisation and leads to formation of a texture close to basal type, with enhanced basal polar density round the normal to the rolling plane. The cross-rolling single component texture $\{10\bar{1}4\}$ is characterised by high polar density, ~ 9 , and a relatively small divergence between the basal and rolling planes⁹. The loss of a hot rolling prismatic component texture during subsequent cross-rolling of the sheet is associated with a deficiency of easy shear systems, $\{10\bar{1}0\}\langle 11\bar{2}0 \rangle$ and $(0001)\langle 11\bar{2}0 \rangle$, for prismatic orientated grains; the only possible additional mechanism of deformation by twinning transforms the initial prismatic component into an orientation close to the basal type (Fig.6).

Studies of the technical treatment of $(\alpha + \beta)$ titanium alloys by pressure give little possibility for the strengthening of divergent basal texture, due to suppression of prismatic texture development. Further improvement of basal texture in $(\alpha + \beta)$ alloys is possible by means of micro-alloying, to aid the splitting of basal dislocations, by enhancing the probability of basal slip and by growth of a central type basal component texture.

CONCLUSIONS

- (1) During deformation of titanium alloys in the temperature range for existence of the α -phase, the deformation mechanism is the main condition determining texture development. Alloying the α - or β -phases with elemental stabilisers changes the tendency to dislocation splitting along the basal plane and stimulates the development of a basal or double-basal texture.
- (2) During deformation of titanium alloys in the temperature range for existence of the β -phase, the texture of the transformed α -phase is determined both by the deformation texture and/or β -phase recrystallisation, and by the orientation of the externally applied stresses with respect to the internal stresses arising during inverse transformation.
- (3) Variation of deformation conditions - of temperature, degree and rate of reduction, and also direction of principal deformations - permits wide modifications of the texture in α and $(\alpha + \beta)$ titanium alloys and confers given orientations, useful in defining the working conditions for components made from these alloys.

Improvement of texture permits a 45% increase in the structural strength of α -alloys. In thermally strengthened $(\alpha + \beta)$ titanium alloys with reduced plasticity, the development of given orientations permits the avoidance of texture weakening and makes use of the effect of thermal strengthening under rigid biaxial stress conditions⁹.

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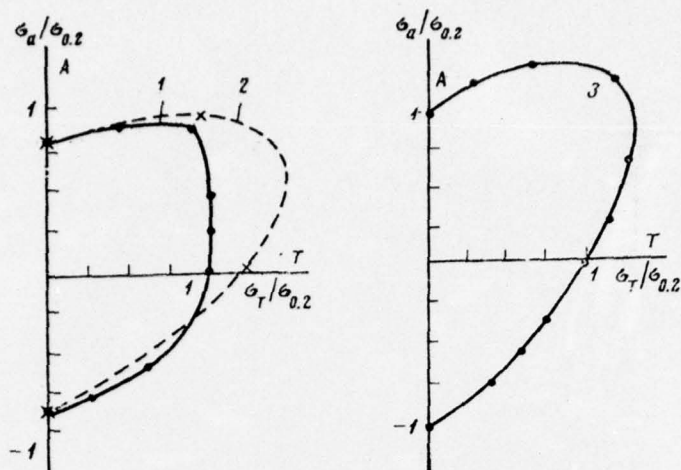


Fig.1 Yield curves for tubes of Zircaloy-2 and -4 alloys

Key:

- 1 = alloy 4 with hydrides and $\langle 0001 \rangle$ texture in the tangential direction
- 2 = alloy 2 with $\langle 0001 \rangle$ texture in the tangential direction
- 3 = alloy 4 with hydrides and double basal texture
- A = axial direction
- T = tangential direction

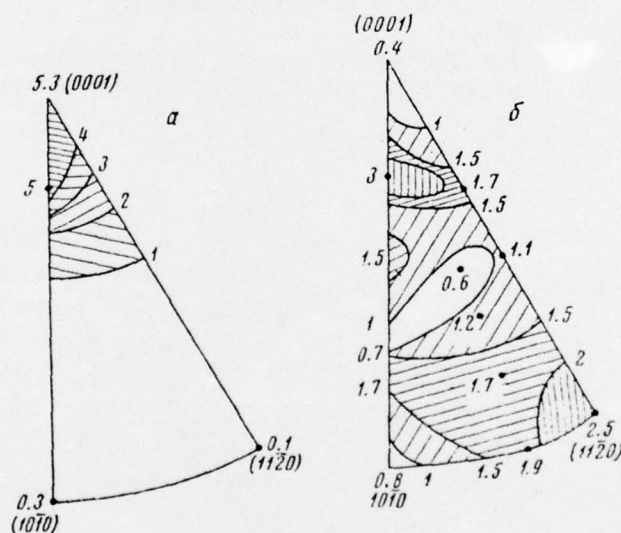


Fig.2 Inverse polar diagrams, normal to the plane of the sheet, VT-5-1 alloy

Key:

- a = rolling at 800°C with 70% reduction
- b = rolling at 800°C with 70% reduction, followed by reheating for 1h at 1100°C

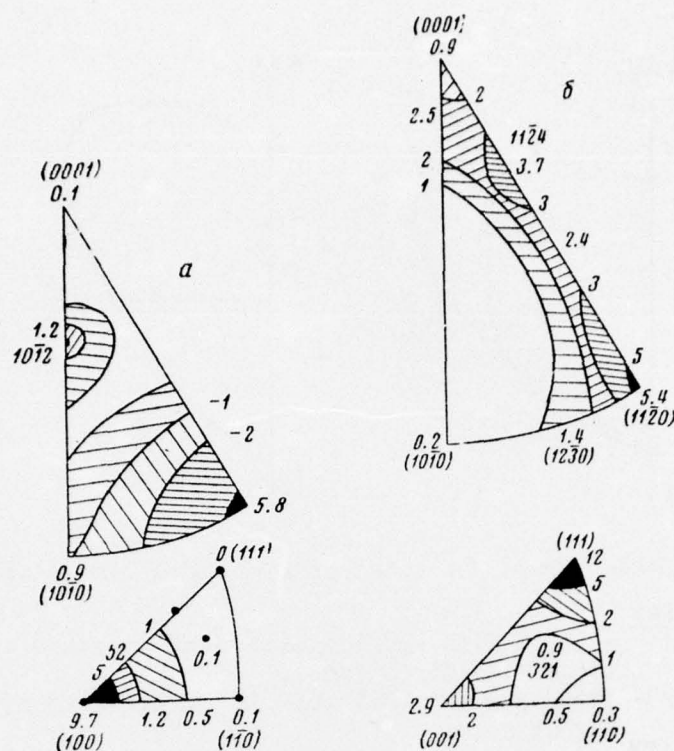


Fig.3 Inverse polar diagrams for α - and β -phases

Key:

a = for the radial direction in stamping of VT-3-1 alloy

b = for the normal to the rolling plane, hot-rolled VT-14 alloy sheet

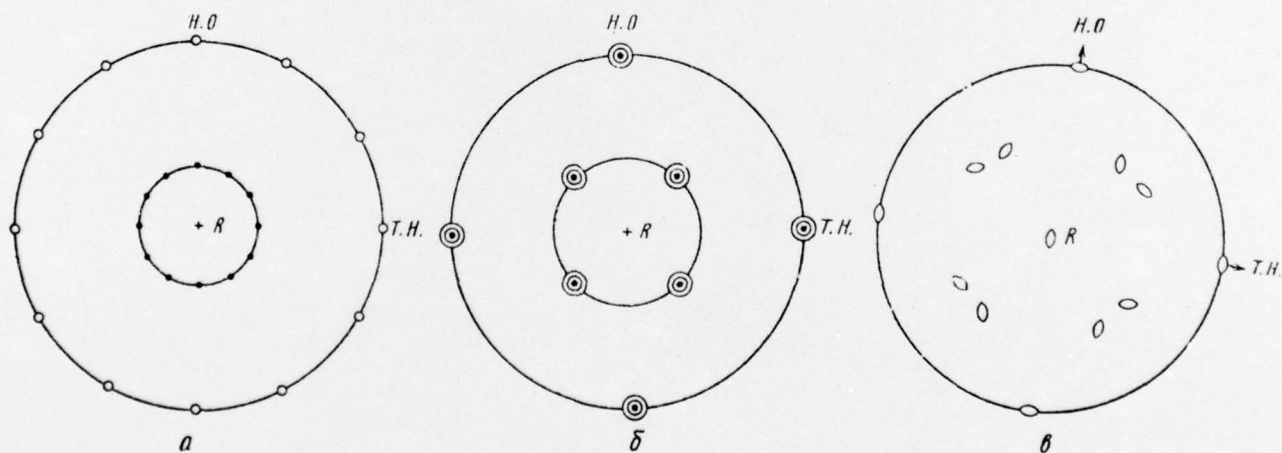


Fig.4 Theoretical polar diagrams $(0001)_\alpha || (110)_\beta$, constructed for initial β -phase texture

Key:

H.O. = direction of forming

T.H. = tangential direction in cup

a = $\{111\}\langle 1\bar{1}0 \rangle$

b = $\{100\}\langle 011 \rangle$

c = $\{110\}\langle 1\bar{1}0 \rangle$ and $\{110\}\langle 001 \rangle$

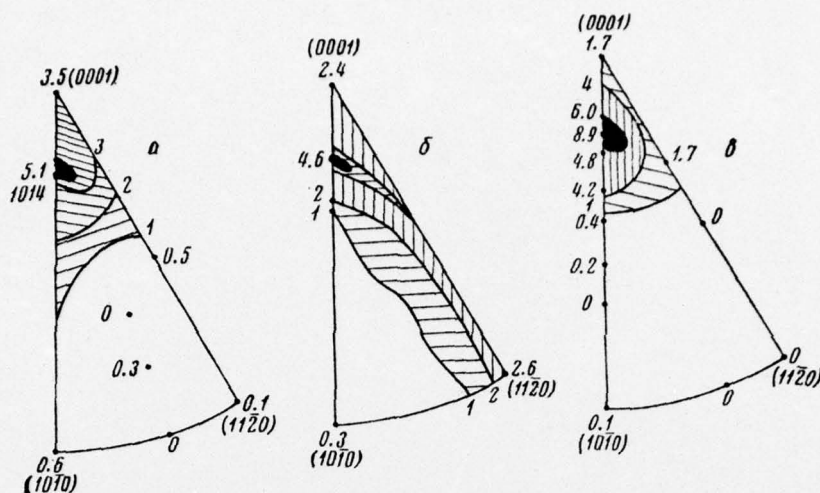


Fig.5 Inverse polar diagrams normal to the rolling plane for VT-14 alloy

Key:

- a = direct rolling at 800°C with 70% reduction
- b = direct rolling at 700°C with 70% reduction
- c = cross-rolling at 700°C with 70% reduction

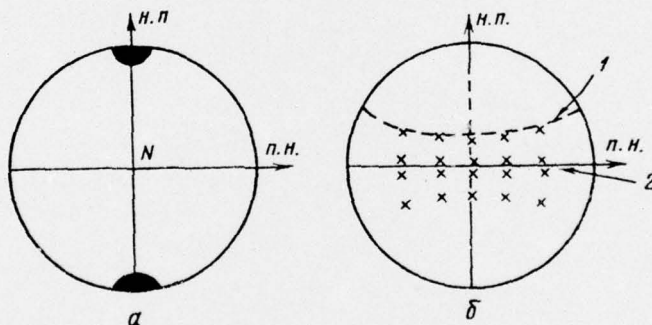


Fig.6 Transformation of the prismatic texture component into a divergent basal texture by deformation twinning during rolling in a lateral direction: polar diagrams (0001)

Key:

- a = initial orientation
- b = final orientation after twinning
- 1 = along $\{11\bar{2}2\}$ planes
- 2 = along $\{10\bar{1}2\}$ planes